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Deformation of Ni–Al at high temperature

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Nickel-rich Ni–Al was deformed in the temperature range 720 K to 1200 K. The deformation mechanism was found to be independent of the composition. An anomalous behaviour of the activation volume is described in the stoichiometric compound and is attributed to the easy activation of [1 1 0] slip in the temperature range 850 K to 950 K.

1. Introduction

The deformation behaviour of nickel-rich $Ni_{1+x}Al_{1-x}$ has been mostly studied and interpreted below 900 K, where the hardening due to non-stoichiometry is high [1–4].

Above 900 K, where the yield stress is strongly thermo-activated, the deformation mechanism is still unclear. Many authors agree with the value of the activation energy (around 2.7 eV) associated with a diffusion process [5–7] but the following problems still require clarification.

(a) The role of the "easy" and "hard" slip systems. The "easy" slip direction at low temperature is the ⟨1 0 0⟩. At high temperature other slip vectors are expected to occur because of the good plasticity of the compound. These "hard" slip directions are ⟨1 1 0⟩ or ⟨1 1 1⟩; they have been actually observed in extruded polycrystals [8] and in single crystals compressed parallel to ⟨0 0 1⟩ (hard orientation) [9]. The ⟨1 1 0⟩ and ⟨1 1 1⟩ dislocations were interpreted as the results of interactions between appropriate ⟨1 0 0⟩ segments rather than as a result of a true slip system operating from the beginning. The first objective of this work was to check if the "hard" slip needed the previous activation of the [1 0 0] systems.

(b) The influence of composition. We have tried to determine the activation parameters of the deformation as a function of composition.

2. Experimental details

Compositions of $Ni_{1+x}Al_{1-x}$ in the range $0 \leqslant x \leqslant 0.09$ were used. The samples were in the form of either thin sheets deformed in a bending machine [4] or small rods compressed at a constant rate.

In the bending tests, the activation volume was measured through instantaneous strain-rate changes. In compression tests, the activation volumes were calculated from stress relaxation experiments.

Our results did not fit with a power-law relation, $\dot{\epsilon} \propto \sigma^n$ between strain rate, $\dot{\epsilon}$, and effective stress, σ; on the contrary, they could be satisfactorily explained by an exponential relationship

$$\dot{\epsilon} = \dot{\epsilon}_0 \exp -\frac{\Delta H - \sigma^* V}{kT} \tag{1}$$

where ΔH is the activation energy, V the activation volume, σ^* the effective stress, k the Boltzmann factor, T the absolute temperature, and $\dot{\epsilon}_0$ is the pre-exponential factor, as described by Giui and Pratt [10], the stress decrement, $\Delta\sigma$, during the relaxation increases with time, t, according to

$$\Delta\sigma \propto kT/V \log (t + c) \tag{2}$$

where c is a constant which can be fitted with the experimental results. The activation parameters could not be measured by the usual way of studying the steady-state creep rate as a function of stress and temperature. Indeed, some compression creep tests were made, which did not exhibit a well-defined steady state. The activation energy was, thus, calculated from the activation volumes and from the temperature dependence of yield stress which had been previously determined [3, 4].

The temperature range used in this work was 720 to 1200 K.

3. Microstructure after deformation

The deformation microstructure was found to be the same in all the range of temperatures and compositions.

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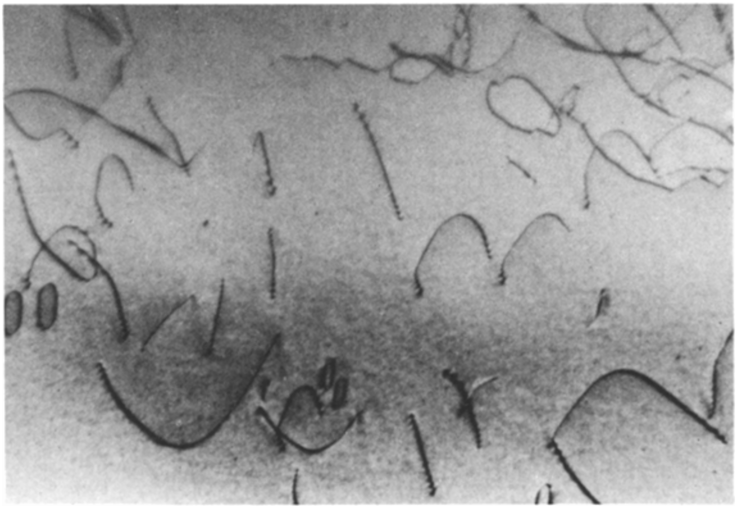

Figure 1 Large ⟨1 0 0⟩ loops after deformation at 870 K.

Most dislocations had a ⟨1 0 0⟩ Burgers vector; numerous large loops were observed, which had a strong prismatic character (Fig. 1); these loops could not originate from combined glide and climb movements.

In slightly deformed grains, evidence of ⟨1 1 0⟩ slip was found, an example is shown in Fig. 2: two sets of dislocations with vectors [1 0 1] and [0 1 1] are seen and, at their intersections, extended nodes are limited by ⟨1 0 0⟩ dislocations. The complete identification of the Burgers vectors is given in Fig. 2b. In that case, the ⟨1 1 0⟩ dislocations correspond to actual slip systems, which do not presume a previous activation of ⟨1 0 0⟩ dislocations. The large size of the triangular nodes limited by [1 0 0] dislocations indicates that the gain of energy due to the reaction [1 0 0] + [0 1 0] → [1 1 0] is small although is high enough to stabilize the [1 1 0] dislocation. This is consistent with the calculations of Lloyd and Loretto [8]. In highly strained specimens, numerous [1 1 0] and [1 1 1] segments were found but, as reported in [9], they were generally the results of interactions between appropriate [1 0 0] dislocation.

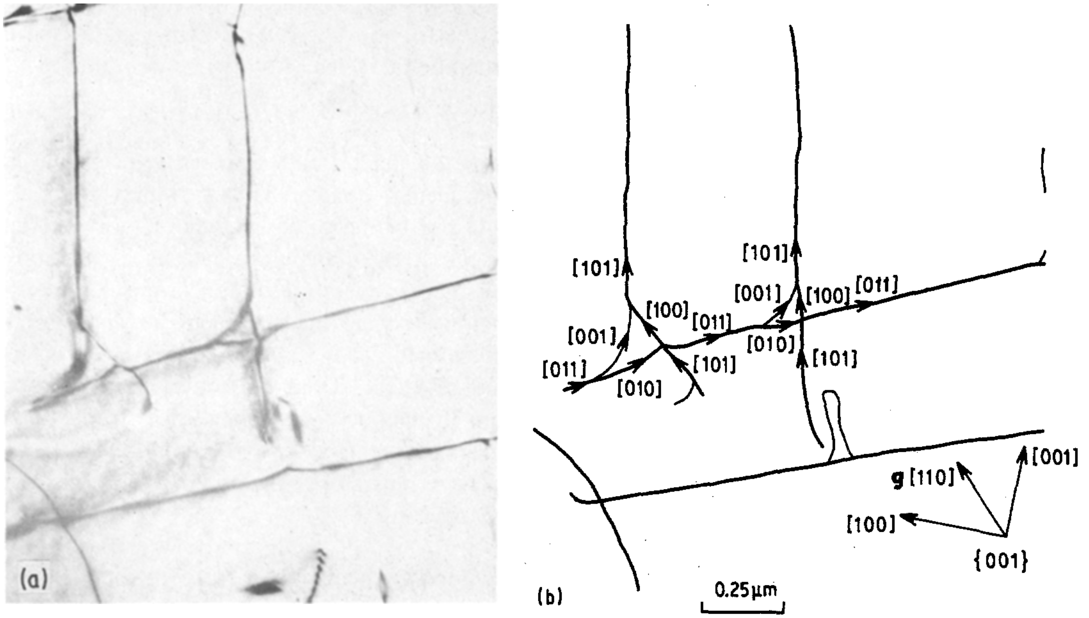

Figure 2 ⟨1 1 0⟩ dislocations ($T = 950$ K, $x = 0.04$ and $\epsilon = 10^{-2}$). (a) Bright-field image, $\mathbf{g} = (1\,1\,0)$ and (b) identification of the Burgers vectors.

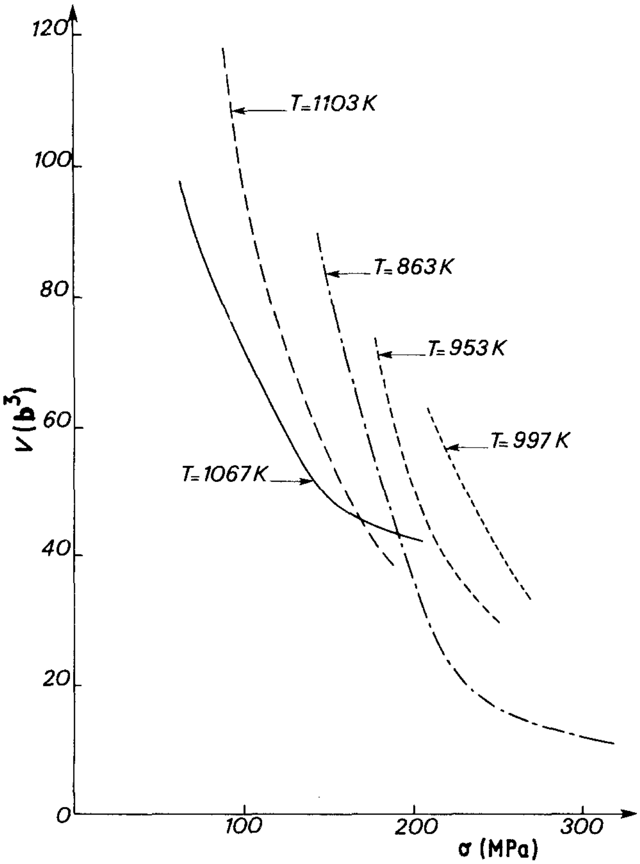

Figure 3 Activation volumes in the stoichiometric compound.

4. Mechanical tests

In this work a distinction will be made between the different behaviours of stoichiometric and non-stoichiometric compounds.

4.1. Stoichiometric compound

Fig. 3 describes the variation of the activation volume, V, with stress and temperature.

V behaves in a rather anomalous way since, for a given stress, it goes through a maximum value around 990 K (see Fig. 4). This anomaly can be qualitatively understood if one refers to the results of Pascoe and Newey [1]. According to these authors, the flow stress for hard slip becomes equal to or smaller than that for easy slip between 850 K and 1050 K (see Fig. 5a). The temperature dependence of the activation volumes is sketched in Fig. 5b.

Our results can thus be explained as follows: below 990 K (the maximum of the $V(T)$ curve), the deformation is controlled by the movement of the "hard" dislocations and $\mathrm{d}V/\mathrm{d}T > 0$ according to Fig. 5b; from the previous paragraph, these "hard" dislocations have the [1 1 0] Burgers vector. Above 990 K, [1 0 0] slip is again easier than [1 1 0] slip and V decreases.

There is thus, apparently, a domain where [1 1 0] slip is easier than [1 0 0] slip; it seems to

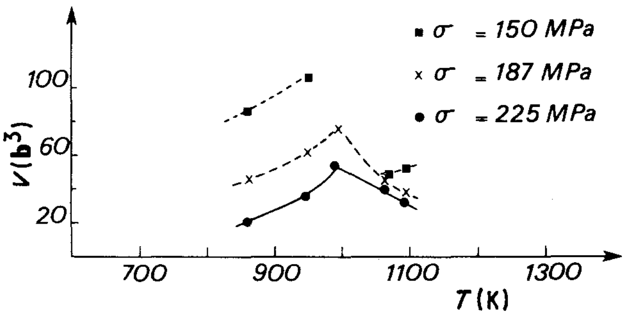

Figure 4 $V(T)$ plots for several stress levels.

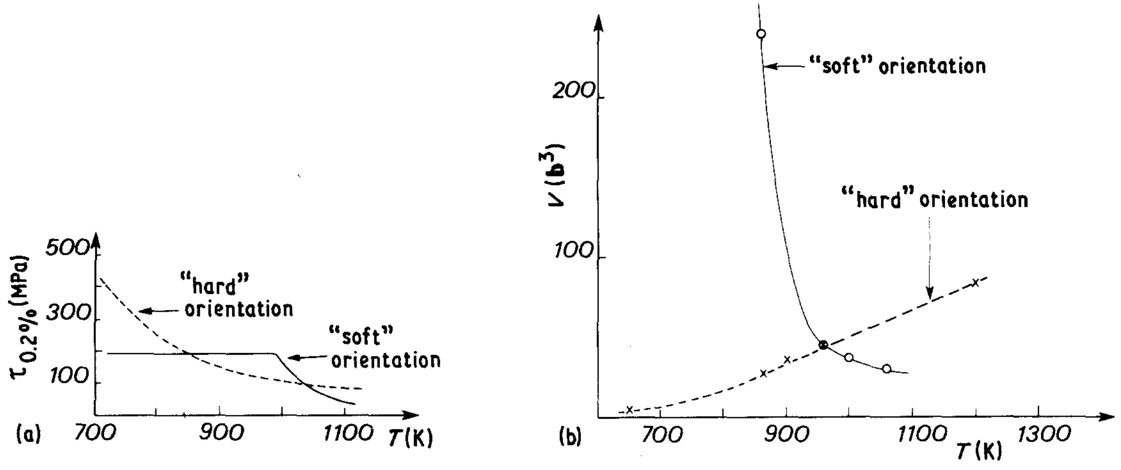

Figure 5 (a) Flow stress for easy or hard slip systems (according to [1]). (b) Activation volume for the easy and hard slip systems (according to [1]).

apply only to the first stage of deformation since the anomaly in $V(T)$ is more pronounced for low stresses.

4.2. Non-stoichiometric compounds

No anomaly in the $V(T)$ behaviour was found as soon as $x \geqslant 0.02$. The activation volume is, then, independent of both temperature and composition (see Fig. 6). All the results for $0.02 < x < 0.06$ and $870\,\text{K} < T < 1170\,\text{K}$ can be fitted with a relation

$$\sigma V/b^3 = 5000 \simeq \mu/20, \qquad (3)$$

where μ is the shear modulus in the [1 0 0] direction and $\sigma V/b^3$ is in MPa.

The activation energy calculated from V and from the $\sigma(T)$ curves published elsewhere [2, 4] was approximately 3 ± 0.5 eV for all compositions. The precision of ΔH determination is poor, but this value is similar to those reported by various authors [5–7].

The observed microstructure (large prismatic loops) and the value of ΔH suggest a mechanism that is controlled by climb and independent of the composition.

Since no anomaly was found in the $V(T)$ curve,

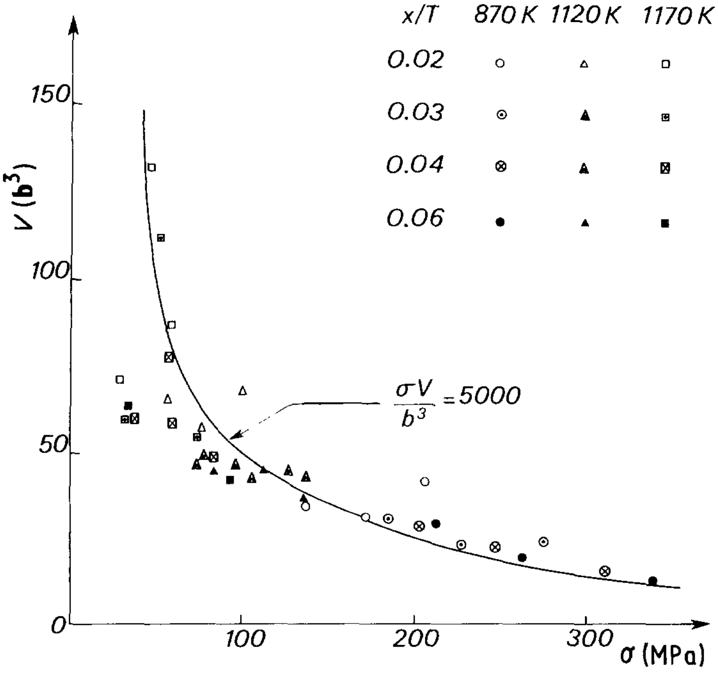

Figure 6 Activation volumes in nickel-rich $Ni_{1+x}Al_{1-x}$.

the present authors consider that the deformation rate is always controlled by the [1 0 0] dislocations; in the non-stoichiometric compounds, the stress for [1 1 0] slip is thus always larger than for [1 0 0] slip; the observed [1 1 0] dislocations are thus the result either of [1 0 0] interactions or of stress concentrations in some areas.

5. Conclusion

Above 900 K, the deformation mechanism is independent of the concentration in all the nickel-rich compounds; it is controlled by the climb and glide of [1 0 0] dislocations. In the close vicinity of stoichiometry [1 1 0] becomes, in the temperature range 850 to 990 K, easier than [1 0 0] slip at least for small strains.

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